(19) Europäisches Patentamt
European Patent Office
Office européen des brevets

(11) Publication number: **0 462 779 A2**

(12) EUROPEAN PATENT APPLICATION

(21) Application number: **91305456.5**

(51) Int. Cl.5: **C21D 8/02**, C21D 9/02

(22) Date of filing: **17.06.91**

(30) Priority: **19.06.90 JP 158790/90**
28.01.91 JP 25076/91
28.01.91 JP 25077/91

(43) Date of publication of application:
27.12.91 Bulletin 91/52

(84) Designated Contracting States:
DE FR GB

(71) Applicant: **NISSHIN STEEL CO., LTD.**
4-1 Marunouchi 3-chome
Chiyoda-ku Tokyo 100(JP)

(72) Inventor: **Suzaki, Tsunetoshi, c/o Tekkokenkyusho**
Nisshin Steel Co., Ltd., 11-1, Showa-cho
Kure-shi, Hiroshima-ken(JP)
Inventor: **Iwao, Tomoyoshi, c/o Tekkokenkyusho**
Nisshin Steel Co., Ltd., 11-1, Showa-cho
Kure-shi, Hiroshima-ken(JP)
Inventor: **Tanaka, Teruo, Nisshin Steel Co., Ltd.**
4-1 Marunouchi 3-chome, Chiyoda-ku
Tokyo(JP)
Inventor: **Yamada, Toshiro, c/o Tekkokenkyusho**
Nisshin Steel Co., Ltd., 5, Ishizunishi-machi
Sakai-shi, Osaka(JP)

(74) Representative: **BATCHELLOR, KIRK & CO.**
2 Pear Tree Court Farringdon Road
London EC1R 0DS(GB)

(54) Method of making steel useful in springs.

(57) Steel material containing by weight from 0.4 % to 0.8 % carbon, from 0.5 % to 2.5 % silicon, from 0.3 % to 2.0 % manganese, from 0.1 % to 1.5 % chromium, and from 0.1 % to 0.5 % molybdenum is hot-rolled to form a strip or plate. The hot-rolled plate or strip is annealed and cold-rolled at a rolling reduction between 10 % and 80 %. The cold-rolled strip or plate is heated at a temperature above Ac3 critical point for a time sufficient to austenitize carbide and annealed.

Steel springs with attractive properties for use in motor vehicle clutches can be produced from this steel material.

EP 0 462 779 A2

The present invention relates to a method of making steel which is useful in springs, such as the diaphragm spring component in clutches for motor vehicles.

In recent years, the environmental temperature of the spring used in a machine has increased with increase of the output power of the machine. For example, clutch torque developed in motor vehicle clutches can increase due to increase of the engine power of the vehicle, particularly in four-wheel drive vehicles. As a result, the environmental temperature of the clutch increases up to 250-350°C from the ordinary maximum of 150°C found in conventional motor vehicle clutch springs.

The diaphragm spring is made of carbon tool steel such as SK5 (Japanese Industrial Standard). However, springs of carbon steel can relax quickly, in that they may become inoperative when the temperature thereof reaches or exceeds the high environmental temperature of 150°C.

It is known that if the silicon content of steel is increased, endurance of the spring, that is a property of the spring resisting heat without settling, increases. However, conventional steels which include a large silicon content can be liable to relax at a high temperature.

An object of the present invention is to provide a method of making steel having a property resisting high temperature, whereby a spring made of the steel is more resistant to relaxation at relatively high temperatures.

Another object of the present invention is to provide steel which may be quickly quenched at a low temperature, thereby preventing significant endurance reduction in the steel.

It has now been surprisingly found that steel having better or even excellent endurance against high temperature relaxation of springs made therefrom could be made by properly controlling solid solution and precipitation of carbide in steel which includes carbon (C), silicon (Si), manganese (Mn), chromium (Cr), molybdenum (Mo) and optionally others.

According to the present invention, the method for making steel comprises hot-rolling steel material consisting essentially by weight of from 0.4 % to 0.8 % carbon, from 0.5 % to 2.5 % silicon, from 0.3 % to 2.0 % manganese, from 0.1 % to 1.5 % chromium, from 0.1 % to 0.5 % molybdenum, and the remainder iron and inevitable impurities to form a plate, annealing the hot-rolled plate, cold-rolling the annealed hot-rolled plate at rolling reduction of 10 % to 80 %, annealing the cold-rolled plate at a temperature below Ac1 critical point, heating the annealed cold-rolled plate at a temperature above Ac3 critical point for a time sufficient to austenitize carbide, cooling the heated cold-rolled plate at a speed higher than a lower critical cooling speed, heating the cooled cold-rolled plate for a time necessary for precipitating carbide and then cooling it to a room temperature.

The lower critical cooling speed is a speed above which the austenite is fully transformed to the martensite.

In the last heating process, molybdenum carbide is finely precipitated, thereby preventing the dislocation migration which causes the relaxation of the spring at high temperature. The heating is preferably performed at a temperature between 450°C and 600°C for a time sufficient to precipitate the carbide.

The silicon content and chromium content are preferably selected so as to satisfy the equation:

$$-7 \leq 4 \times Si(\%) - 10 \times Cr(\%) \leq 5.$$

The heating of the cooled cold-rolled plate is preferably performed so as to provide an annealed hardness between HV400 and HV550.

Furthermore, the annealing of the cold-rolled plate is preferably performed at a temperature between 550°C and 730°C, thereby providing carbide having an average grain diameter less than 2µm.

In order that the invention may be readily appreciated and carried into effect, embodiments thereof will now be described by way of example only, and with reference to the accompanying drawing.

The figure is a graph showing relationship between heating temperature and hardness of steel.

DETAILED DESCRIPTION OF THE PREFERRED EMBODIMENTS

Quantity of each component in the steel, preparation conditions and factors for their numerical limitation are now described.

(Carbon)

Carbon is effective in increasing the strength of steel. In order to obtain a strength necessary for the the spring, carbon content of 0.4 % or more by weight must be included. However, if carbon is included in

excess of 0.8 %, quenching crack and reduction of toughness of steel occur. Therefore the carbon is included in the range 0.4 % to 0.8 % by weight.

(Silicon)

In the method of the present invention, the material is tempered at high temperature. Silicon is added to prevent the strength from reducing due to the high temperature tempering. It is necessary to add silicon of 0.5 % or more by weight. If the silicon content exceeds 2.5 %, internal oxidation and decarburization which are unfavourable to the spring occur, and graphitization is enhanced in the hot rolling and annealing.

(Manganese)

Manganese is effective in deoxidizing steel and in increasing the hardenability of the steel, if the manganese is included at 0.3 % or more by weight. If the manganese content exceeds 2.0 %, toughness of the steel reduces significantly after quenching and tempering.

(Chromium)

Chromium acts to restrict the graphitization and the internal oxidation which are enhanced by silicon, and is effective in increasing the hardenability as is effected by manganese, if chromium is included at 0.1 % or more by weight. If the chromium content exceeds 1.5 %, toughness of the steel reduces after quenching and tempering.

Moreover, Si content and Cr content are most preferably determined to satisfy the following equation, thereby preventing decarburization and graphitization.

$$-7 \leq 4 \times Si(\%) - 10 \times Cr(\%) \leq 5 \qquad (1)$$

(Molybdenum)

The molybdenum included in the steel of the present invention forms carbide in the steel after cold rolling and annealing. The carbon becomes solid solution in austenite when the steel is heated over the Ac3 critical point. Consequently, the austenite is transformed into martensite after quenching, and carbide separates finely upon tempering at high temperature, thereby significantly increasing endurance whilst withstanding against relaxation. In order to obtain such an effect, it is necessary to include molybdenum of at least 0.1 % but no more than 0.5 % by weight. If the molybdenum content exceeds 0.5 %, a large amount of carbide remains without becoming solid solution in austenite when the steel is heated above the Ac3 critical point.

(Vanadium, Niobium)- optional but preferred component(s)

vanadium and niobium if included in the steel of the present invention become carbide after the cold rolling and annealing thereof. Remaining vanadium and niobium without becoming solid solution in austenite act to prevent austenite grain from growing. On the other hand, solid solution of vanadium and niobium in austenite are in solid solution in martensite when quenching, and precipitate finely as carbide when tempering, thereby enhancing endurance whilst withstanding against relaxation. In order to attain these effects, vanadium and/or niobium each in an amount of 0.05 % or more are necessary. If the individual content of either or both exceeds 0.5 %, the quantity of undissolved carbide in austenite increases when the steel is heated above Ac3 point, thereby reducing fatigue strength of the steel.

(Aluminium)- a component which is preferably substantially absent.

The steel spring is fatigued by repeated bending or twisting. Existence of hard inclusions such as aluminum aggravates this fatigue. To reduce the influence of any such hard inclusion, the aluminium content of the steel is preferably less than 0.020 weight percent.

Preferred manufacturing conditions are now described, for steel plate or strip.

In the cold rolling, when rolling reduction is smaller than 10 %, the grain size of carbide becomes

coarse when annealed below the critical point Ac1. Consequently, a long time is required for transforming the carbide to austenite when heated above the Ac3 critical point, which causes an increase in decarburization and hence spring characteristic is deteriorated. When the rolling reduction is larger than 80 %, work hardening due to the cold rolling is remarkably increased, causing deformation such as edge crack. Therefore, an upper limit is 80 %.

If the annealing after the cold rolling is performed at a temperature above 730°C (Ac1 critical point), spheroidized grain of carbide becomes coarse. Consequently, it takes a long time to transform the carbide to austenite, resulting in increase of decarburization causing deterioration of spring characteristic. Therefore, the annealing after the cold rolling is carried out at a temperature below the Ac1 point. If the annealing temperature is lower than 550°C, the hardness increases, so that the formability of the material reduces. Therefore, the annealing temperature is between 550°C and 730°C.

If the average grain diameter of carbide after the annealing is less than 2µm, carbide is easily dissolved austenite at quenching. Therefore, it is necessary to maintain the average grain diameter of carbide to a value smaller than 2µm for effectively performing the quenching.

In order to increase the strength of the steel made by the cold rolling and annealing to a value necessary for the spring, the strip is heated at a temperature higher than the critical point Ac3 for a time sufficient for austenitizing the spheroidal carbide, after which it is cooled at a speed higher than a lower critical cooling speed, namely quenching. Thereafter, the strip is heated at a temperature between 450°C and 600°C for a time to precipitate fine carbide and cooled to a room temperature (that is tempering). At the quenching, the parent material is austenitized by heating it over the Ac3 point, and then carbon and other elements are dissolved to martensite by cooling at a speed higher than the lower critical cooling speed. By tempering the material at a temperature higher than 450°C, carbide of Mo, V and Nb is finely precipitated from the martensite, thereby increasing the endurance withstanding against the relaxation. If the tempering is carried out at a higher temperature than 600°C, a carbide of Mo, V and Nb becomes coarse which can not prevent the dislocation migration. In addition, the strength of the steel largely reduces. Therefore, the tempering is performed at a temperature below 600°C.

Example 1

Table 1 shows contents of steels. In the table, A to F are steels of the present invention, and G to L are comparative steels.

Each of the steels A to F is made into a hot-rolled plate of 3.5 mmt by ordinary hot rolling and then the plate is annealed and cold rolled at a rolling reduction between 5 % and 90 %. Thereafter, the steel is annealed at 700°C below the Ac1 point for 10 hours, and is soaked at 900°C above the Ac3 point for a period necessary to provide remaining carbide ratio below 1 % by weight. Thereafter, the steel is quenched into oil.

Table 2 shows results of tests for edge crack and depth of decarburization. When the rolling reduction exceeds 80 %, edge crack occurs. If the rolling reduction is smaller than 10 %, carbide becomes coarse. Consequently, it takes a long time to dissolve carbide into austenite, so that the depth of decarburization increases significantly.

TABLE 1

Steel	C	Si	Mn	P	S	Cr	Mo	V	Nb
A	0.58	1.65	0.73	0.011	0.008	0.95	0.21	Tr	Tr
B	0.61	1.69	0.73	0.012	0.006	0.97	0.22	0.22	Tr
C	0.59	1.66	0.73	0.010	0.007	0.96	0.22	Tr	0.19
D	0.62	1.65	0.75	0.012	0.007	0.95	0.21	0.21	0.11
E	0.64	1.71	0.82	0.016	0.009	0.72	0.31	0.24	0.21
F	0.72	2.21	1.52	0.015	0.008	0.74	0.42	0.31	0.22
G	0.32	1.62	0.74	0.014	0.009	0.95	0.23	Tr	Tr
H	0.92	1.64	0.77	0.015	0.008	0.92	0.22	0.01	Tr
I	0.62	0.21	0.81	0.013	0.006	0.93	0.21	Tr	Tr
J	0.59	1.61	0.21	0.013	0.008	0.92	0.21	0.22	Tr
K	0.62	1.72	0.80	0.012	0.006	0.02	0.21	0.02	Tr
L	0.62	1.59	0.71	0.011	0.008	0.92	0.01	Tr	0.01

TABLE 2

Steel	Rolling Reduction (%)	Edge Crack	Annealing Temp. (°C)	Quenching Temp. (°C)	Depth of Decarburization (μ m)
A	5	NOT FOUND	700	900	6.1
	15	NOT FOUND	700	900	3.1
	30	NOT FOUND	700	900	2.1
	70	NOT FOUND	700	900	1.2
	90	FOUND	—	—	—
F	5	NOT FOUND	700	900	7.2
	15	NOT FOUND	700	900	3.4
	40	NOT FOUND	700	900	2.2
	70	NOT FOUND	700	900	1.3
	90	FOUND	—	—	—

Example 2

Each of the steels A to F is made into a hot-rolled plate of 3.5 mmt by ordinary hot rolling and annealed and cold rolled at rolling reduction of 35 % to form a cold rolled plate of 2.3 mm. Thereafter, the steel is annealed once at 700° C for 10 hours, and is heated at a temperature between 850° C and 900° C for 10 minutes. Thereafter, the steel is quenched into oil and tempered at a temperature between 420° C and 630° C for 30 minutes.

A relaxation test was performed in order to estimate endurance against relaxation. The test was carried out at 350° C, initial 1.0 % strain, holding time of 12 hours. Load reduction after the test was regarded as relaxation rate.

Table 3 shows the result of the relaxation test. Since comparative example G is smaller than the present invention in carbon content, comparative example 1 is smaller in silicon content, comparative example J is in manganese content, and K is in chromium content, each of these steels has low strength so that the relaxation rate thereof is high. Although the comparative example H has a large carbon content, the relaxation rate is not largely reduced. Since the comparative example L has no molybdenum, the carbide of which is effective to increase the endurance, relaxation rate is very high. Although each of comparative

examples A', D' and F' has the same ingredient content as the present invention, the tempering temperature is out of the range of the present invention. Consequently, the relaxation rate is not largely reduced.

To the contrary each steel according to the present invention has a very low relaxation rate comparing with the comparative examples, which means that the steel has a high endurance withstanding against the relaxation.

TABLE 3

	Steel	Quenching Temp. (°C)	Tempering Temp. (°C)	Hardness (HV)	Relaxation Rate (%)
present invention	A	900	480	496	16.2
		900	520	475	15.1
		900	560	452	14.4
		850	520	462	15.7
	B	900	560	470	13.5
	C	900	520	479	15.7
	D	900	480	513	14.2
		900	520	492	13.4
		900	560	468	12.6
		850	560	452	13.2
	E	900	560	453	14.1
	F	900	580	472	11.1
comparative example	G	900	520	348	40.2
	H	900	560	473	20.1
	I	880	520	394	25.2
	J	900	520	442	18.2
	K	900	560	421	21.7
	L	880	520	427	32.5
comparative example	A'	900	420	567	21.5
		900	630	413	19.2
	D'	900	420	591	20.1
		900	630	426	18.1
	F'	900	420	625	19.5
		900	630	513	17.3

Example 3

Embodiments A to G in Table 4 are steels of the present invention and H to L are comparative steels.

TABLE 4

(wt.%)

Steel	C	Si	Mn	P	S	Cr	Mo	V	Nb	Al	SICR value
A	0.70	1.99	0.80	0.011	0.004	0.52	0.19	Tr.	Tr.	0.007	2.76
B	0.71	2.02	0.42	0.012	0.003	0.90	0.23	Tr.	Tr.	0.006	-0.92
C	0.62	1.51	0.51	0.010	0.003	0.42	0.22	Tr.	0.20	0.008	1.84
D	0.61	1.51	0.73	0.011	0.004	0.51	0.31	Tr.	Tr.	0.007	0.94
E	0.49	1.12	1.28	0.011	0.004	0.92	0.21	0.22	Tr.	0.005	-4.72
F	0.60	2.11	0.55	0.012	0.005	0.32	0.22	Tr.	Tr.	0.006	4.24
G	0.59	1.55	0.70	0.011	0.004	0.51	0.31	0.21	0.11	0.008	1.10
H	0.71	1.52	0.51	0.012	0.004	1.35	0.22	Tr.	Tr.	0.009	-7.42
I	0.71	2.02	0.81	0.012	0.003	0.52	0.20	Tr.	Tr.	0.028	2.88
J	0.60	2.25	0.70	0.012	0.004	0.15	0.21	Tr.	0.01	0.006	7.50
K	0.82	0.21	0.36	0.012	0.004	0.01	Tr.	Tr.	Tr.	0.007	0.74
L	0.70	2.00	0.78	0.013	0.003	0.51	Tr.	Tr.	Tr.	0.008	2.97
M	0.72	0.21	0.82	0.010	0.003	0.52	0.21	Tr.	Tr.	0.008	-4.36

SICR value=4×Si (%) − 10×Cr (%)

Each of the steels in the table was hot-rolled to provide a hot-rolled plate having a thickness of 3.5 mmt, and then annealing the hot-rolled plate. The plate was cold rolled at rolling reduction of 35 % to prepare a cold-rolled plate of 2.3 mmt thickness. The cold-rolled plate was annealed at a temperature between 650° C and 750° C for 10 hours to provide a test piece. Hardenability test was performed in such a manner that the test piece was rapidly heated to 850° C at the rate of 140C°/sec, heated from 850° C to a test temperature between 900° C and 1100° C at the rate of 30C°/sec, and then rapidly cooled immediately after the heating without taking a holding time. The hardenability was estimated by the hardness of the test piece after the quenching. Results of the test are shown in the attached figure.

As is seen from the graph, the test piece A having ingredient contents according to the present invention has an average grain diameter of carbide less than 2µm when annealed at 650° C and 700° C. Even if the test piece A is heated to the lowest temperature 900° C, the hardness becomes the higher

value. However, if it is annealed at 750°C so that the average grain diameter exceeds 2μm, the hardness does not reach the highest value unless the quenching temperature is elevated up to 950°C.

The comparative example H has a SICR value of -7.42 out of the range of the present invention. Consequently a large amount of chromium remains in carbide after the annealing. Accordingly, the steel must be heated up to 1000°C in order to obtain the higher hardness, although the average grain diameter is smaller than 2μm.

From the comparisons it will be seen that the carbide is rapidly dissolved into austenite at a lower temperature in accordance with the present invention.

Fatigue test and relaxation test piece are estimated as follows. The cold-rolled plate having 2.3 mm thickness is annealed at 680°C for 10 hours, and then heated at 900°C and quenched. Thereafter, a plurality of the plates are tempered at various temperatures for 30 minutes.

The fatigue test was performed in alternating plane bending fatigue. The result of the test is shown in Table 5.

TABLE 5

Steel	Tempering Temp. (℃)	Hardness (HV)	Test Temp. (℃)	Fatigue Strength (kgf/mm²)
A	580	454	25	52
			250	51
	540	490	25	57
G	550	452	25	53
			250	51
	500	505	25	56
I	580	448	25	47
			250	46
J	590	452	25	43
			250	41

From the table, it will be seen that although the steel A of the present invention has a hardness approximately equal to the comparative example I, the steel A is superior to the comparative example I in fatigue strength. This is caused by the fact that the aluminum content of the steel A is less than 0.020 weight percent, which means hard inclusion causing fatigue fracture is small. The steel G has the same fatigue characteristic as steel A.

The comparative steel J has a small Cr content compared with Si content, so that SICR value is 7.50 out of the range of the present invention, producing graphite at annealing. In addition, since a long time was required for austenitization, decarburization increased. As a result, the fatigue characteristic is inferior to the steels A and G.

The endurance withstanding against relaxation was estimated by the relaxation test. Table 6 shows test results.

TABLE 6

Steel	Quenching Temp. (℃)	Tempering Temp. (℃)	Hardness (HV)	Relaxation Rate (%)
A	900	580	454	12.4
		540	490	12.2
B	900	580	461	13.1
		540	494	13.1
C	900	520	471	16.3
D	900	540	469	12.0
		510	496	11.9
E	900	520	435	17.3
F	900	560	455	12.5
G	900	550	452	11.5
		500	505	11.3
K	850	430	450	36.7
	850	400	475	38.5
L	900	550	459	31.5
M	880	490	465	25.2

Example 4

In table 7, A to G are steels of the present invention, H to L are comparative steels.

TABLE 7

Steel	C	Si	Mn	P	S	Cr	Mo	V	Nb	Al
A	0.58	1.65	0.73	0.011	0.005	0.95	0.21	Tr.	Tr.	0.007
B	0.61	1.69	0.73	0.012	0.004	0.97	0.22	0.22	Tr.	0.006
C	0.59	1.66	0.73	0.010	0.003	0.96	0.22	Tr.	0.19	0.006
D	0.63	1.12	0.71	0.012	0.002	0.32	0.32	0.20	0.10	0.008
E	0.71	2.01	0.52	0.016	0.004	0.51	0.23	Tr.	Tr.	0.009
F	0.72	2.21	1.52	0.012	0.004	1.21	0.21	Tr.	Tr.	0.007
G	0.70	1.01	0.53	0.011	0.002	0.53	0.21	0.12	Tr.	0.007
H	0.32	1.62	0.74	0.014	0.003	0.95	0.23	Tr.	Tr.	0.007
I	0.70	2.03	0.71	0.011	0.004	0.53	0.21	Tr.	Tr.	0.031
J	0.62	0.21	0.81	0.013	0.002	0.93	0.21	Tr.	Tr.	0.006
K	0.62	1.59	0.71	0.011	0.003	0.92	0.01	Tr.	0.01	0.008
L	0.85	0.25	0.40	0.011	0.003	0.02	Tr.	Tr.	Tr.	0.008

Each of the steels in the table was hot rolled to provide a hot-rolled plate having a thickness of 3.5 mmt, and the hot-rolled plate was then annealed. The plate was cold rolled at rolling reduction of 35 % to prepare a cold-rolled plate of 2.3 mmt thickness. The cold-rolled plate was annealed at 680 °C for 10 hours, and then heated at a temperature between 850 °C and 900 °C for 10 minutes and quenched into oil. All plates were tempered at various temperatures for 30 minutes.

The fatigue test was performed in alternating plane bending fatigue. The result of the test is shown in Table 8.

TABLE 8

Steel	Quenching Temp. (℃)	Tempering Temp. (℃)	Hardness (H V)	Testing Temp. (℃)	Fatigue Strength (kgf/mm²)	
A	900	540	463	25	51	I
		480	508	25	58	I
		410	573	25	47	II
E	900	580	452	25	51	I
				250	50	I
		510	516	25	57	I
		440	578	25	47	II
I	900	580	448	25	46	II
				250	44	II

note: I represents present invention steels, while II represents comparative steels.

From the table, it will be seen that although the steel E of the present invention has a hardness approximately equal to the comparative example I, the steel E is superior to the comparative example I in fatigue strength due to the lower aluminium content.

Even if the steel components are within the range of the present invention, the fatigue strength may reduce if the annealed hardness exceeds HV550.

Endurance against settling was estimated by the relaxation test. Test temperature was 350° C, initial strain 1.0 %, and holding time 12 hours. Table 9 shows the test results.

TABLE 9

Steel	Quenching Temp. (℃)	Tempering Temp. (℃)	Hardness (HV)	Relaxation Rate (%)	Notes
A	900	480	496	16.2	present invention
		520	475	15.1	
		560	452	14.4	
	850	520	462	15.7	
B	900	560	470	13.5	
C	900	520	470	15.7	
D	900	440	509	16.1	
		480	491	15.9	
		520	465	15.2	
	850	520	449	16.2	
E	900	580	453	12.5	
F	900	580	472	11.1	
G	900	520	446	13.6	
H	900	520	348	40.2	comparative example I
J	880	520	394	25.2	
K	880	520	427	32.5	
L	850	430	450	36.7	
A	900	650	391	22.1	comparative example II
D	900	620	382	23.1	
E	900	660	392	23.2	
G	900	620	378	25.1	

Comparative steels H and J have small C content and Si content, and hence they have high relaxation rates, respectively. Since comparative steel K has no Mo, it has a high relaxation rate. Even if each of steels A, D, E and G include components within the present invention, the relaxation rate is not significantly reduced if the tempering temperature increases and hardness is lower than an annealed hardness of HV400, as shown in comparative examples II.

While presently preferred embodiments of the present invention have been shown and described, it is to be understood that such disclosure is only for illustration and that various changes and modifications may be made without departing from the scope of the invention as set forth in the appended claims.

Claims

1. A method for making steel comprising:

hot-rolling steel material consisting essentially by weight of from 0.4 % to 0.8 % carbon, from 0.5 % to 2.5 % silicon, from 0.3 % to 2.0 % manganese, from 0.1 % to 1.5 % chromium, from 0.1 % to 0.5

% molybdenum,the remainder substantially iron and inevitable impurities to form a strip or plate;
annealing the hot-rolled strip or plate;
cold-rolling the annealed hot-rolled strip or plate at a rolling reduction between 10 % and 80 %;
annealing the cold-rolled strip or plate at a temperature below Ac1 critical point;
heating the annealed cold-rolled strip or plate at a temperature above Ac3 critical point for a time sufficient to austenitize carbide;
cooling the heated cold-rolled strip or plate;
heating the cooled cold-rolled strip or plate for a time necessary for precipitating carbide and then cooling it to a room temperature.

2. A method according to claim 1 wherein the steel material further comprises
0.05 % to 0.5 % by weight of vanadium and/or 0.05 % to 0.5 % by weight of niobium.

3. A method according to claim 1 or 2 wherein the steel material further includes no more than 0.020 % by weight of aluminium.

4. A method according to any preceding claim claim wherein heating of the cooled cold-rolled stripp or plate is performed at a temperature of 450 °C to 600 °C.

5. A method according to any preceding claim wherein the silicon and chromium content of the steel are selected to satisfy the equation:

$$-7 \leq 4 \times Si(\%) - 10 \times Cr(\%) \leq 5 \qquad (1)$$

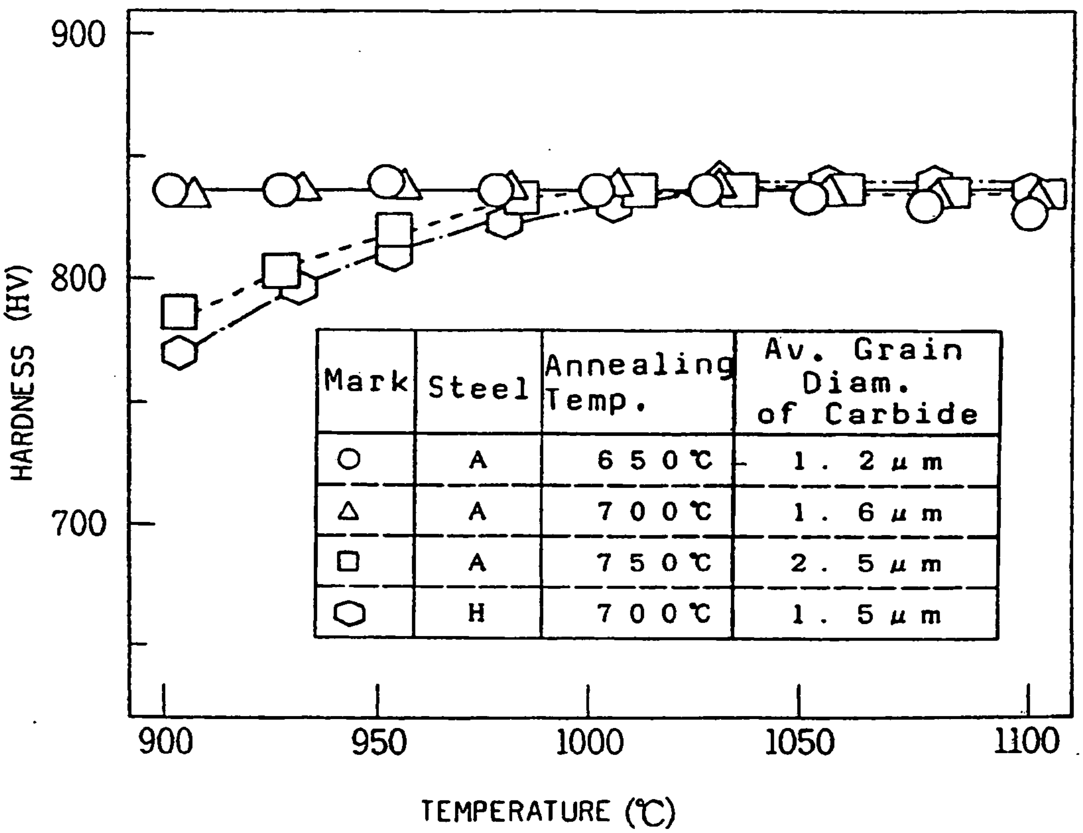
6. A method according to any preceding claim wherein the cooling of the heated cold-rolled strip or plate is performed at a speed higher than a lower critical cooling speed.

7. A method according to any preceding claim wherein heating of the cooled cold-rolled strip or plate is performed to provide an annealed hardness of HV400 to HV550.

8. A method according to any preceding claim wherein annealing of the cold-rolled strip or plate performed at a temperature of 550 °C to 730 °C, such as to provide carbide having an average grain diameter less than 2μm.

9. A spring constructed of a steel consisting essentially by weight of 0.4 % to 0.8 % carbon, 0.5 % to 2.5 % silicon, 0.3 % to 2.0 % manganese, 0.1 % to 1.5 % chromium, from 0.1 % to 0.5 % molybdenum, the remainder substantially iron and inevitable impurities, optionally including 0.05% to 0.5% vandadium and/or 0.05% to 0.5% niobium and/or no more than 0.020% of aluminium.

10. Use of a steel material obtained by a method as claimed in any one of claims 1 to 8 in the production of a diaphragm spring for a motor vehicle clutch.



Mark	Steel	Annealing Temp.	Av. Grain Diam. of Carbide
○	A	650℃	1.2 μm
△	A	700℃	1.6 μm
□	A	750℃	2.5 μm
⬡	H	700℃	1.5 μm